REPORT DOCUMENTATION PAGE

Form Approved OMB No. 0704-0188

1. AGENCY USE ONLY (*Leave Blank*)

2. REPORT DATE
November 2, 1995

3. REPORT TYPE AND DATES COVERED
Final Report June 1, 1993-May 31, 1995

4. TITLE AND SUBTITLE
Structural Characterization of Epitaxial Layers for Infrared Detectors

5. FUNDING NUMBERS
~~F49620-92-J-0285~~
61103D
3484/S3

6. AUTHOR(S)
D. W. Greve
Professor of Electrical and Computer Engineering

7. PERFORMING ORGANIZATION NAME(S) and ADDRESS(ES)
Carnegie Mellon University
Schenley Park
Pittsburgh, PA 15213

AFOSR-TR-95
0712

9. SPONSORING/MONITORING AGENCY NAME(S) AND ADDRESS(ES)
AFOSR/NE
Building 410
Bolling AFB, DC 20332-6448
Interim Contract Monitor:
Dr. Michael Prairie

10. SPONSORING/MONITORING AGENCY REPORT NUMBER
F49620-92-J-0285

11. SUPPLEMENTARY NOTES

12a. DISTRIBUTION/AVAILABILITY STATEMENT
Approved for public release; distribution unlimited.

12b. DISTRIBUTION CODE

13. ABSTRACT (*Maximum 200 words*)

UHV/ CVD is a growth technique highly suitable for deposition of Ge_xSi_{1-x} heterostructures for long- wavelength infrared detectors. We have used transmission electron microscopy to determine favorable conditions for the growth of these structures. Multiple quantum well structures can be grown with excellent quality without any evidence of nonplanar growth, while heterojunction internal photoemission structures incorporated thicker Ge_xSi_{1-x} layers do exhibit nonplanar growth. A modest decrease in growth temperature to 550 °C is sufficient to solve the problem.

14. SUBJECT TERMS
Epitaxial, Germanium, Silicon, Infrared

15. NUMBER OF PAGES

16. PRICE CODE

17. SECURITY CLASSIFICATION OF REPORT	18. SECURITY CLASSIFICATION OF THIS PAGE	19. SECURITY CLASSIFICATION OF ABSTRACT	20. LIMITATION OF ABSTRACT
UNCLASSIFIED	UNCLASSIFIED	UNCLASSIFIED	Unlimited

NSN 7540-01-280-5500

Standard Form 298 (Rev. 2-89)
Prescribed by ANSI Std 239-18
298-102

Structural Characterization of Epitaxial Layers for Infrared Detectors

Final Report
AFOSR Contract F49620-92-J-0285

June 1, 1993- May 31, 1995

D.W. Greve
Department of Electrical and Computer Engineering
Carnegie Mellon University
Pittsburgh, PA

Research Objectives

This contract is an Augmentation Award for Science and Engineering Research Training (AASERT) grant associated with primary contract concerning the growth of epitaxial heterostructures for infrared detectors. The specific objective of this contract is to perform structural characterization of germanium- silicon heterostructures grown by UHV/ CVD.

This grant provided support for one graduate student and a summer undergraduate student.

Project Status and Personnel Changes

A no- cost extension for this contract was requested but not approved. Consequently no additional progress has occurred since the previous report. The student expected to work on this project will not be supported after May, 1996.

Status of the Research Effort

The parent grant (AFOSR F49620-92-J-0155 and its successor F49620-95-1-0161) supports research on the development of germanium- silicon epitaxial structures for infrared detectors. In the following, I will detail the role of research supported under this contract. Two detector structures were under investigation: multiple quantum well structures and heterojunction internal photoemission detectors (Fig. 1 (a) and (b), respectively). The objective of our work was to develop a detector structure which could be incorporated in silicon- based focal plane arrays for the far infrared (8- 12 μm) region. Such detectors must (1) offer BLIP performance with a 300 K background; and (2) be sensitive to normally incident illumination.

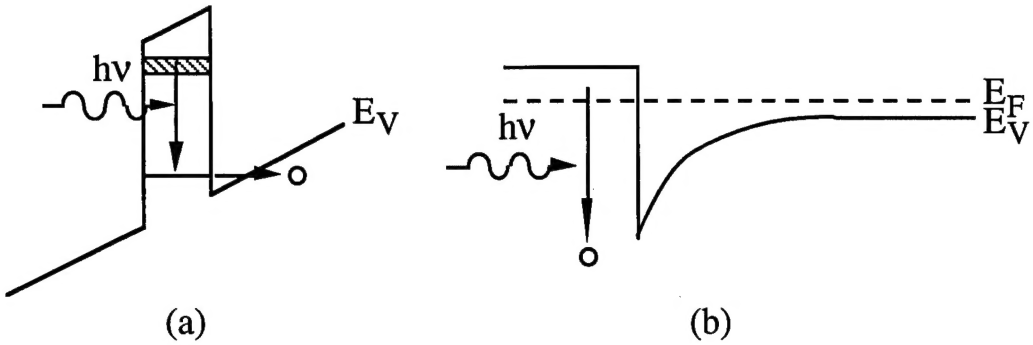

Fig. 1. Detector structures: (a) quantum well and (b) HIP detector.

One motivation for the quantum well photodetector is the possibility of particularly strong absorption when the first excited state is aligned near the silicon valence band. On the other hand, intersubband transitions (for example, the heavy hole ground state, HH0, to the first heavy hole excited state, HH1) are forbidden transitions in simple effective mass theory for normally incident radiation. However, a more complete treatment of the valence band suggests that these transitions will occur. In 1992, approximately at the time we started our work, experimental reports included only MBE- grown material and appeared to be contradictory; with one group reported observable normal incidence transitions in the 8- 12 μm region [1] while another reported normal incidence transistions only at shorter wavelengths [2]. Our objective was to grow well- characterized samples by UHV/ CVD; to determine the normal incidence absorption properties; and if these results were promising, to fabricate and characterize detectors.

TEM studies conducted under this contract played a major role in determining the general quality of our epitaxial layers and also in particular the quality of multiple quantum well samples. Figure 2 shows a TEM cross section of such a sample (10- 60Å $Ge_{0.28}Si_{0.72}$ wells doped with 4 x 10^{18} cm^{-3} boron; 298 Å spacer;). The top and bottom interfaces of the quantum wells are abrupt and planar and no defects are visible. Measurements on many similar samples with doped wells showed that only the free- carrier absorption process was detectable for normal incidence illumination [3]. A similar conclusion was also drawn independently by Robbins et al. [4]. These results are in contradiction to those of People et al. [1]. As no strong intersubband transitions were observed under the conditions of interest, we have chosen to emphasize the other detector type, the HIP detector.

The HIP detector (Fig. 1b) requires thicker layers (≈ 150 Å) and here our TEM studies showed that initially the films were not satisfactory. We observed undulation formation in layers grown at 600 °C- our standard growth temperature- which were of the desired composition and thickness. Undulation formation is a consequence of elastic relaxation which occurs by surface migration of adatoms. This is illustrated vividly in Fig. 3 which shows the two $Ge_{0.32}Si_{0.68}$ layers of the same composition and growth time, with a pause after growth of the top layer before capping with silicon. Reducing the growth temperature modestly to 550 °C permits growth of planar films. The results of this study were presented at the Electronic Materials Conference in June, 1994 [5] and a journal article is in preparation [6]. Parts of this work were also presented at another conference [7] and later published [8]. The growth procedure developed as part of this

work has now been used to fabricate HIP detectors which have excellent performance, with C1 values comparable to those of PtSi detectors in the 3- 5 μm regime [9,10].

Finally, I will note the work performed by summer undergraduate students. During the summer of 1993 P.K. Banh studied the deposition of SiO_2 layers from the liquid phase for possible use as a low- temperature deposited passivation layer on germanium- silicon heterojunctions. This was motivated by difficulties with edge leakage currents that were observed at that time. The resulting layers were not superior to plasma- deposited layers which were being used.

C. Weber was employed as an undergraduate research assistant during the summer of 1994. He implemented models of the Ge_xSi_{1-x} growth rates using a user- friendly software package. The model will be useful for the prediction of growth rates at temperatures and flow conditions which have not yet been measured. In addition, C. Weber performed some software development for computer control of our electrical measuring instruments.

References

1. R. People, J.C. Bean, C.G. Bethea, S.K. Sputz, and L.J. Peticolas, *Appl. Phys. Lett.* **61**, 1122 (1992).
2. R.P.G. Karunasiri, J.S. Park, and K.L. Wang, *Appl. Phys. Lett.* **59**, 2588 (1991).
3. "Infrared Absorption in Ge_xSi_{1-x}/ Si Quantum Wells," R. Misra, D.W. Greve, and T.E. Schlesinger, (presented at the 1994 Electronic Materials Conference, Boulder, CO).
4. D.J. Robbins, M.B. Stanaway, W.Y. Leong, R.T. Carline, and N.T. Gordon, *Appl. Phys. Lett.* **66**, 1512 (1995).
5. "Growth and Characterization of Ge_xSi_{1-x} Films by Ultra High Vacuum Chemical Vapor Deposition for Infrared Detector Applications," S. Vyas, R. Strong, T.J. Knight, D.W. Greve, and T.E. Schlesinger, (presented at the 1994 Electronic Materials Conference, Boulder, CO).
6. "Morphology of boron- doped and undoped epitaxial Ge_xSi_{1-x} layers grown on (001) Si," S.M. Vyas, R. Strong, V. Balakrishna, D.W. Greve and S. Mahajan (in preparation).
7. "Growth of Epitaxial Ge_xSi_{1-x} for Infrared Dectectors by UHV/ CVD," S. Vyas, D.W. Greve, T. Knight, R. Strong, and S. Mahajan, (presented at the 4th European Vacuum Conference, Uppsala, Sweden, June, 1994).
8. "Growth of Epitaxial Ge_xSi_{1-x} for Infrared Detectors by UHV/ CVD, S. Vyas, D.W. Greve, T.J. Knight, R.M. Strong, and S. Mahajan, (to appear in *Vacuum*).
9. "Ge_xSi_{1-x} Heterojunction Internal Photoemission Structures by Ultra- High Vacuum Chemical Vapor Deposition," R. Strong, D.W. Greve, T.E. Schlesinger, M.M. Weeks, and P. Pellegrini, (to appear in *Proc. MRS Symposium C on Strained Layer Epitaxy- Materials, Processing, and Device Applications*).
10. "Ge_xSi_{1-x} infrared detectors grown by ultrahigh- vacuum chemical- vapor deposition for forcal- plane arrays," R. Strong, D.W. Greve, T.E. Schlesinger, M. Weeks, and P. Pellegrini, (extended abstract in CLEO 1995 Technical Digest, pp. 357- 358).

Publications

(This section lists cumulatively all publications, conference proceedings, and presentations specifically relating to this grant. A comprehensive list of publications under the parent grant has been prepared separately).

Journal Articles

"Growth of Epitaxial Ge_xSi_{1-x} for Infrared Detectors by UHV/ CVD, S. Vyas, D.W. Greve, T.J. Knight, R.M. Strong, and S. Mahajan, *Vacuum* **8-10**, 1065 (1995).*

"Morphology of boron- doped and undoped epitaxial Ge_xSi_{1-x} layers grown on (001) Si," S.M. Vyas, R. Strong, V. Balakrishna, D.W. Greve and S. Mahajan (in preparation).

Conference Proceedings

"Characterization of SiO_2 Films Deposited from the Liquid Phase at 40°C," D.W. Greve and P.K. Banh, presented at the *Active Matrix Liquid Crystal Displays Symposium*, (Bethlehem, PA, October, 1993).

Conference Presentations

"Heterojunction Infrared Diodes Using Ge_xSi_{1-x} Films Grown by Ultra High Vacuum Chemical Vapor Deposition," R. Strong, T.J. Knight, S.M. Vyas, D.W. Greve, and T.E. Schlesinger, (presented at 1994 American Physical Society Spring Meeting).

"Growth of Epitaxial Ge_xSi_{1-x} for Infrared Dectectors by UHV/ CVD," S. Vyas, D.W. Greve, T. Knight, R. Strong, and S. Mahajan, (presented at the 4th European Vacuum Conference, Uppsala, Sweden, June, 1994).

"Growth and Characterization of Ge_xSi_{1-x} Films by Ultra High Vacuum Chemical Vapor Deposition for Infrared Detector Applications, S. Vyas, R. Strong, T.J. Knight, D.W. Greve, and T.E. Schlesinger, (presented at the 1994 Electronic Materials Conference, Boulder, CO).

"Characterization of SiO_2 Films Deposited from the Liquid Phase at 40°C," D.W. Greve and P.K. Banh, *Proceedings of the Active Matrix Liquid Crystal Displays Symposium*, pp. 68- 71 (Bethlehem, PA, October, 1993).

* A copy of this publication has been appended to this report.

Personnel

Faculty

D.W. Greve, Professor ECE, principal investigator

S. Mahajan, Professor MSE, co- advisor of S. Vyas

Graduate Students

Sanjay Vyas (Ph.D. student MSE, supported by AASERT grant, US citizen; until September, 1994).

Summer Undergraduate Student

Cory Weber
(Junior undergraduate at Carnegie Mellon and ROTC member; supported by AASERT grant; US citizen).

Interactions

Wright Laboratory, WPAFB
O. Manasreh- FTIR
W. Mitchel- spectrally resolved photoconductivity

Air Force Institute of Technology
R.L. Hengehold- FTIR and photoluminescence

Rome Laboratory
P. Pellegrini- optical characterization of detectors

Other- degrees awarded

S. Vyas, M.S. in Materials Science, 1994.

Inventions/ Patent Disclosures

(none).

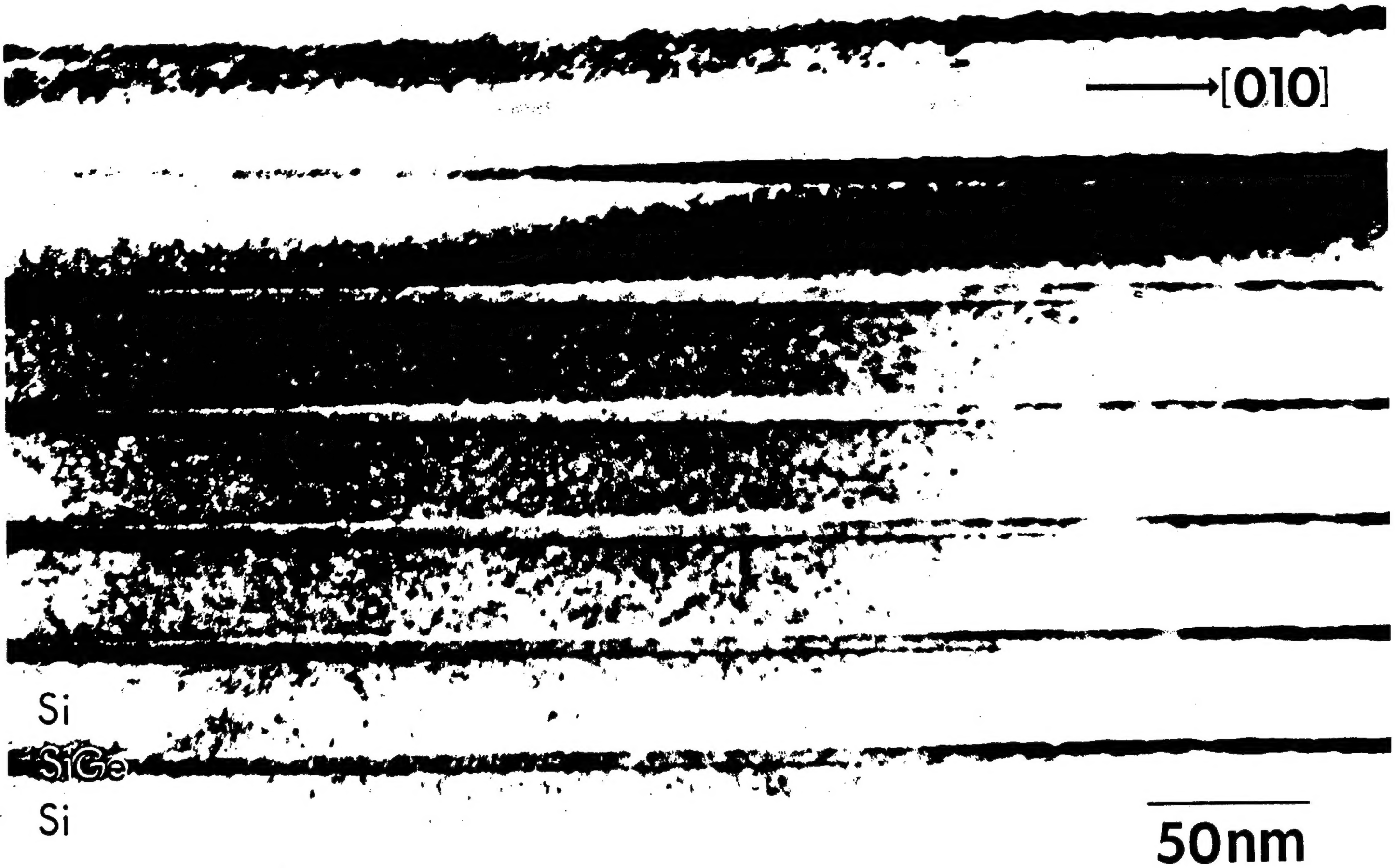

Figure 2.

Figure 3.

Pergamon

Vacuum/volume 46/numbers 8–10/pages 1065 to 1069/1995
Elsevier Science Ltd
Printed in Great Britain
0042–207X/95 $9.50+.00

0042–207X(95)00108–5

Growth of epitaxial Ge_xSi_{1-x} for infrared detectors by UHV/CVD

S Vyas*, D W Greve†, T J Knight†, R M Strong† and S Mahajan* **Department of Materials Science and Engineering; †Department of Electrical and Computer Engineering, Carnegie Mellon University, Pittsburgh, PA 15213, USA*

We report the application of UHV/CVD (ultra-high vacuum chemical vapor deposition) to the growth of heterojunction internal photoemission (HIP) infrared detector structures. The HIP structure is essentially a Schottky barrier junction formed between a degenerately doped germanium–silicon absorbing layer and an undoped or lightly doped silicon layer. This device places stringent demands on the background doping concentration and morphology of the epitaxial layers. C(V) profiling shows that acceptably low doping of the silicon layer is obtained only after a cleanup run. Surface morphology of the Ge_xSi_{1-x} absorbing layer is degraded by the formation of thickness undulations, especially at the large boron concentrations required. By appropriate changes to the growth procedure, we have been able to fabricate working detectors with good electrical characteristics.

Introduction

Infrared focal plane arrays are of interest for various imaging applications in one of the atmospheric transmission 'windows' (3–5 μm or 8–12 μm). In general, monolithic arrays are preferred because of the difficulty of bonding large numbers of separately fabricated detectors and also because of potential reliability problems[1]. Thus detector structures that can be fabricated on the same substrate as is used for silicon readout circuitry are preferred. For the 3–5-μm window, the PtSi–*p* silicon Schottky barrier diode is an excellent detector. In contrast, for the 8–12-μm band, various detector structures have been explored, but none has yet reached a similar level of maturity to the PtSi detector. In this paper, we discuss the fabrication of one of the competing detector designs for the 8–12-μm band, namely, the HIP (heterojunction internal photoemission detector). This detector structure was first fabricated by Lin and Maserjian[2], and the original concept is illustrated in Figure 1(a). In this detector, IR absorption takes place in a heavily doped Ge_xSi_{1-x} epitaxial layer. Holes which gain energy by free carrier absorption may be emitted over the heterojunction barrier to be collected by the silicon substrate.

The detector structure we discuss here is somewhat different in that the heterojunction is formed within the epitaxial layer, rather than at the epitaxial layer–substrate interface. The band diagram of this detector is illustrated in Figure 1(b) and is similar to one reported by Liu *et al.*[3]. In this structure, the heterojunction is formed within the epitaxial layer and prevents barrier-lowering effects associated with boron present at the original growth interface. Boron contamination leading to a 'spike' at the original growth interface is observed in both MBE[4] and low-temperature CVD[5] growth.

The lowest photon energy to which the detector is sensitive is given by:

$$h\nu_{min} = \Delta E_V - (E_{V(GeSi)} - E_F) + \Delta E_{V(bgn)}$$

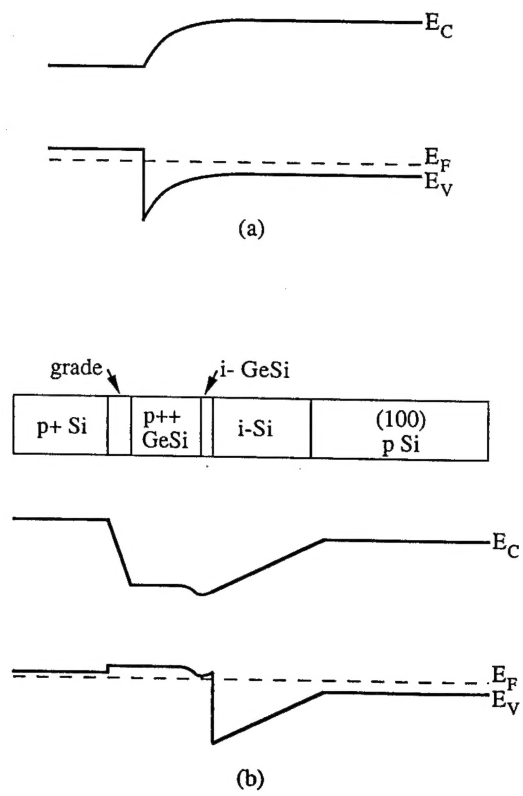

Figure 1. Two heterojunction internal photoemission (HIP) detector structures: (a) structure reported by Lin and Maserjian[2]; (b) structure discussed in this work, similar to that reported by Liu *et al.*[3].

where, for growth on (100) wafers[6], $\Delta E_V \approx 0.84x$ is the valence band offset between silicon and germanium–silicon, $E_{V(GeSi)} - E_F$

increases with the germanium–silicon doping N_A, and $\Delta E_{V(bgn)}$ is the shift in the position of the germanium–silicon band edge due to bandgap narrowing effects. Relatively high doping ($N_A \approx 10^{20}$ cm^{-3}) is necessary for reasonable sensitivity; thus, neglecting bandgap narrowing, a detector with a threshold wavelength of 10 μm has $x \approx 0.30$. Since the actual value of $\Delta E_{V(bgn)}$ is uncertain, it is desirable to explore a range of germanium fractions somewhat smaller than 0.30. In the following sections, we will report the growth and characterization of HIP structures with x and N_A chosen for possible detector operation in the 8–12 μm region.

Growth of Ge_xSi_{1-x} detector structures by UHV/CVD

A wide range of different techniques has been explored for the growth of Ge_xSi_{1-x} heterostructures, including MBE[7,8] and a variety of CVD techniques[9]. We have chosen to use UHV/CVD, a growth technique that utilizes hydride reactants (SiH_4, GeH_4, and B_2H_6). The pressure during growth is about 133×10^{-3} Pa, and the growth temperature is typically 500–600°C. A hot-wall configuration is used, and many wafers are arranged in a wafer boat in a manner similar to that used in oxidation and diffusion furnaces. Uniform growth is possible on many wafers[10] because of the absence of gas-phase reactions (which may result in the formation of highly reactive species) and the small sticking coefficients of the hydride reactants. Abrupt heterojunctions are possible because of the small residence time of reactants, and indeed multiple quantum well structures have been successfully grown[10].

Despite our previous success with growth of multiple quantum well structures, two difficulties were encountered in initial attempts to grow HIP devices. First, high reverse bias currents were observed, which were traced to residual boron doping in the silicon epitaxial layer. Boron doping in this layer has the effect of increasing the electric field and decreasing the thickness of the potential barrier to a point such that tunneling occurs. Secondly, when the layers were grown at 600°C, we observed the formation of thickness undulations of the Ge_xSi_{1-x} layer that were particularly pronounced at the high doping levels required. In the following, we discuss these two difficulties and the changes made to the growth procedure to mitigate them.

Residual boron doping. Figure 2 shows the results of capacitance

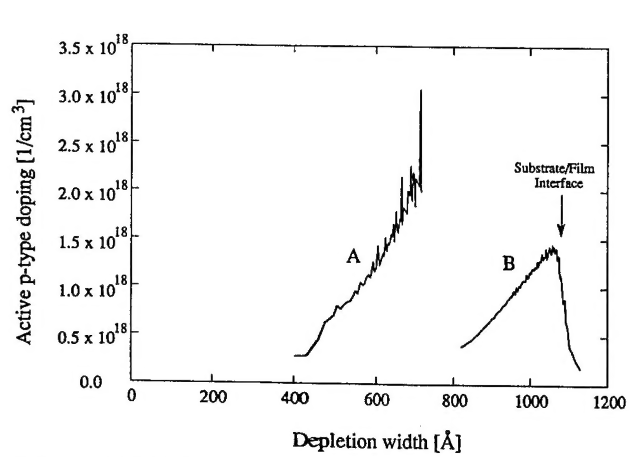

Figure 2. Profile of electrically active p-type dopant as a function of depth for two epitaxial layers. Sample A was 1080 Å thick and was grown after a structure with heavy boron doping; sample B had a similar thickness and was grown after a cleanup run.

profiling of silicon layers grown at 600°C with zero diborane flow. The layers were approximately 1000 Å thick and were grown by using our usual cleaning procedure, which involves an 800°C vacuum prebake immediately before growth of the epitaxial layer. A Schottky diode was then fabricated by depositing and defining an aluminum contact. The growth prior to sample A involved the growth of heavily boron-doped material. Note that the boron concentration decreases as the layer is grown but is sufficiently high to degrade seriously the electrical characteristics of HIP detectors. The presence of significant concentrations of boron cannot be explained by segregation to the growing interface of a boron spike, since this results in much sharper decreases at the growth temperature used[11].

The boron concentration is, however, considerably decreased if a cleanup run (approximately 1000 Å of silicon and germanium–silicon with no diborane flow) is performed. This is illustrated in sample B, which shows a much lower boron concentration in the bulk of the silicon film. The boron concentration is low enough to permit profiling back to the original growth interface, where we observe the boron spike typically seen in MBE and low-temperature CVD growth. The boron concentration at the original growth interface is less than 5×10^{12} boron atoms/cm^2. It is not possible to determine the doping concentration close to the surface accurately; however, from the capacitance measured at zero bias and the known built-in voltage for an aluminum-on-silicon Schottky diode, we can conclude that $N_A < 8 \times 10^{16}$ cm^{-3}.

These results suggest that either diborane or boron can desorb from the walls of the system and redeposit on the growing epitaxial layer. This is contrary to previous studies, which showed that diborane decomposes at 450°C with the release of hydrogen and that the resulting surface boron did not desorb from the surface up to 1000°C[12]. A quadrupole mass spectrometer search for boron-containing species showed no evidence of B_2H_6, BH_3, or B. However, when wafers with a deliberately grown chemical oxide were heated, we did observe signals at $m/q = 26$ and 27, which have been identified as $B^{10}O^{16}$ and $B^{11}O^{16}$, respectively[13]. Such a wafer would have some surface boron oxide from atmospheric contamination[5]. The two peaks had a ratio 1:4.15, which is close to that expected from the boron isotopic abundance. Figure 3 shows the measured intensities of these two peaks as a function of time during heating to the prebake temperature, which demonstrates that these species do desorb during the prebake.

We are thus led to suggest a possible mechanism for the residual boron doping. During a previous growth, diborane decomposes to form surface boron on the walls of the growth chamber. When the wafers are unloaded and new wafers loaded for a subsequent run, enough oxygen and water vapor are introduced to oxidize surface boron. Moderate heating is then sufficient to desorb the boron suboxide that is transported to the silicon surface and reacts again to form surface boron. The cleanup run reduces the residual boron doping by desorbing some of the suboxide and also covering surface boron during growth.

Formation of surface undulations. It has long been recognized that the growth of metastable strained layers is limited by two different processes: the formation of misfit dislocations and the development of non-planar surface morphology. Depending upon growth conditions, one or the other process may constrain device engineering. In the case of the heavily boron-doped Ge_xSi_{1-x} layers used in HIP structures, we expected that these relaxation processes would be less important, since boron has a smaller

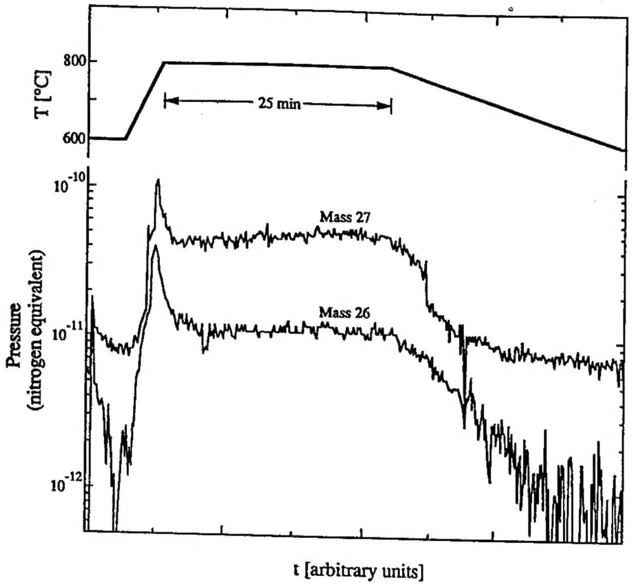

Figure 3. Measured partial pressures for $m/q = 26$ and 27 during prebake of wafers with a chemical oxide. The upper scale indicates the approximate furnace temperature as a function of time.

tetrahedral radius than silicon and thus was expected to reduce the layer strain and the driving force for relaxation.

Our experiments show, however, that this is not the case. In Figure 4, we show cross-sectional TEM photographs from a matrix of samples of two different germanium fractions and

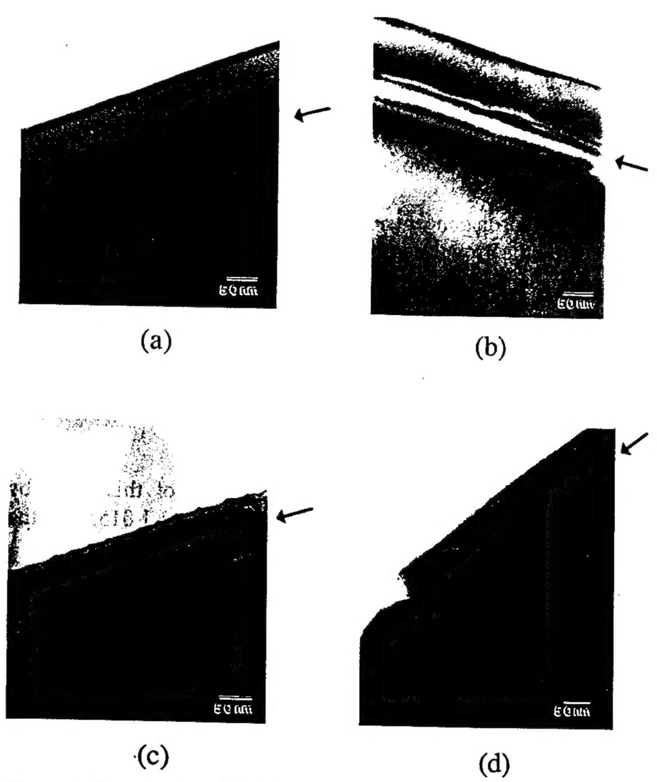

Figure 4. Cross-sectional TEM photographs of Ge_xSi_{1-x} layers grown at 600°C: (a) 380 Å, $x = 0.26$, undoped; (b) 500 Å, $x = 0.26$, $N_A = 8 \times 10^{19}$ cm^{-3}; (c) 300 Å, $x = 0.32$, undoped; and (d) 360 Å, $x = 0.32$, $N_A = 8 \times 10^{19}$ cm^{-3}. All samples consisted of a silicon buffer layer, a Ge_xSi_{1-x} layer, and a silicon cap. The arrows indicate the Ge_xSi_{1-x} layer. Details of the thickness, composition, and doping are in the text.

doping. All samples were grown at 600°C to approximately the same thickness. Before growing the Ge_xSi_{1-x} layer, an approximately 500-Å-thick undoped silicon buffer was grown, and the layers were capped with a thick (approximately 1000-Å) undoped layer of silicon. The germanium fractions were determined by Rutherford backscattering spectrometry (RBS) on thicker samples grown with the same germane and silane flow rates. Doping concentrations were determined by extrapolation of the results of a previously reported study of boron incorporation[14] and are in reasonable agreement with SIMS measurements.

Figure 4(a) shows that the top surface of a 26% undoped layer, 380 Å in thickness, was planar, while a 500-Å layer doped to $N_A = 8 \times 10^{19}$ cm^{-3} (Figure 4(b)) shows some evidence of surface roughness. In Figure 4(c), an undoped 32% layer, 300 Å, thick shows well-developed sinusoidal undulations with a wavelength of about 1100 Å. Finally, the undulations become even more pronounced in a 360-Å doped layer ($N_A = 8 \times 10^{19}$ cm^{-3}) of the same composition. In some cases (notably Figures 4(a), (c), and (d)), the initial growth interface is visible, but in no case are there any visible defects in the active layers. Whereas the layer thicknesses are not exactly comparable, addition of boron causes the undulations to increase proportionately more than the increase in thickness.

Similar undulations have been reported in $In_xGa_{1-x}As$[15] and more recently in undoped Ge_xSi_{1-x} by Pidduck and co-workers[16]. For $x = 0.21$, the amplitude of the undulations was quite small for layers grown at 610°C (amplitude of 15 Å for a total layer thickness of 1180 Å), and pronounced undulations as shown in Figures 4(c) and (d) were observed only in layers of this composition grown at 750°C. Our results show that undulations become quite severe for only moderate increases in germanium fraction and are further aggravated in layers doped to the levels required in HIP devices.

The development of undulations in epitaxial layers and the resulting strains have been studied by Srolovitz[17]. His analysis considered a strained layer with a sinusoidally modulated surface. The formation of undulations reduces the strain energy at the price of an increase in surface energy. Consideration of the variation of chemical potential along the surface showed that undulations with wavelengths greater than a critical value are expected to grow exponentially. For the present case of low-temperature CVD, adatoms are unlikely to desorb once deposited, and thus undulations can grow only through surface diffusion of adatoms. The maximally unstable wavelength is then predicted to be:

$$\lambda_m = (4/3)\pi\gamma / Y\varepsilon^2$$

where Y is the Young's modulus, γ is the surface energy, and ε is the strain. Silicon and germanium crystals are most readily deformed along the $\langle 100 \rangle$ directions (that is, Y_{100} is smallest). Using reported values for γ[18] and for Y_{100} at 600°C[19] yields $\gamma/Y \approx 0.20$ and 0.17 Å for bare germanium and silicon surfaces, respectively. Thus we calculate $\lambda_m \approx 8000$ Å, which is in fair agreement with the observed wavelength of 1100–1600 Å when it is recognized that the surface energy is influenced by step density and hydrogen coverage during growth.

If surface migration is responsible for the growth of surface undulations, then it should be possible to increase the amplitude by a pause after the growth of the Ge_xSi_{1-x} layer. In a test of this hypothesis, an undoped 300-Å $Ge_{0.32}Si_{0.78}$ layer was grown at 600°C followed by growth of a silicon layer approximately 1000 Å in thickness. This was followed by growth of an identical

$Ge_{0.32}Si_{0.78}$ layer followed by a 15-min pause before growth of a silicon cap. TEM showed that the amplitude of the undulations was increased from 60 Å to essentially the entire film thickness during the pause, which confirmed the role of surface migration.

Finally, we consider the impact of undulations on the device characteristics. Srolovitz[17] has shown that an undulated surface is associated with sinusoidal variation in strain along the surface. This strain decays exponentially into the crystal with a decay length comparable with the wavelength of the undulations. Since the band offset is dependent upon the strain state of the interface[20], the result is a sinusoidal variation in ΔE_V, leading to degraded device characteristics. In order to prevent this, we have reduced the growth temperature to 550°C for the devices discussed in the next section. This has the effect of reducing the undulation amplitude in a 250-Å doped film to 20 Å for $x = 0.32$.

Characterization of heterojunction internal photoemission (HIP) structures

We now present the results of recent electrical and optical measurements on HIP photodetector structures. After initiation of growth of a silicon buffer layer at 600°C, the temperature was ramped down to 550°C, resulting in the growth of a buffer layer approximately 1400 Å thick. A 40-Å undoped $Ge_{0.26}Si_{0.74}$ spacer layer was then grown, followed by growth of a heavily doped $Ge_{0.26}Si_{0.74}$ layer about 160 Å thick. This was followed by a 100-Å graded layer from $Ge_{0.26}Si_{0.74}$ to pure silicon. Finally, the temperature was ramped up to 600°C during growth of a 1000-Å-thick heavily doped silicon cap. As discussed above, the $Ge_{0.26}Si_{0.74}$ layer and a portion of the silicon cap were grown at 550°C in order to minimize the formation of undulations. The remainder of the structure was grown at higher temperature to minimize the total growth time. We shall compare below two samples with differently doped absorbing layers (1×10^{19} and 5×10^{19} cm^{-3}).

A mesa process with a plasma SiO_2 passivation layer was used to fabricate detector structures with an area of 6.6×10^{-4} cm^2. Ohmic contacts were formed with sputtered aluminium–copper and covered about 10% of the total device area. Electrical measurements were performed on packaged devices in a cryostat by using an HP 4145 A Semiconductor Parameter Analyzer. Broadband optical sensitivity was determined by illuminating from the front and using a Nernst glower source and InSb and InAs long-pass filters ($h\nu_{cutoff} = 0.17$ and 0.36 eV, respectively).

Figure 5 presents I(V) measurements as a function of temperature for a device with $N_A = 5 \times 10^{19}$ cm^{-3}. The forward-bias ideality factor at 85 K is about 1.05, and, as expected, the current at fixed bias increases with temperature. We have verified that the current in reverse bias is proportional to device area, and, as a result, the reverse current at fixed applied voltage V_A (with $V_A < 0$ for reverse bias) is given by:

$$I(V_A) \approx A^* T^2 \, e^{-q\phi_B(V_A)/kT}$$

where A^* is the effective Richardson constant and $\phi_B(V_A)$ is the Schottky barrier height including barrier lowering. Thus $\phi_B(V_A)$ can be determined by plotting $I(V_A)$ as a function of $1/T$, yielding a barrier height of 0.17 eV and 0.19 eV for the high and low doping concentrations, respectively, for $V_A = -0.3$ V.

Both samples exhibit infrared sensitivity to broadband IR radiation. As expected from the measured value of ϕ_B, no response was observed for the InSb filter. A signal that increased monotonically with reverse bias was easily observed with the InAs filter, however. Again as expected, the signal was greater for the heavily doped sample, for which we estimate a quantum efficiency $\approx 0.5\%$ at $V_A = -1$ V. This is not as high as reported for HIP detectors grown by MBE[21], but this is probably due to the doping concentration, which is about an order of magnitude lower than that reported by Lin *et al.*[21]. For this study, we have kept the boron concentration in the $Ge_{0.26}Si_{0.74}$ active region below 10^{20} cm^{-3} to limit undulation formation during growth. Work is in progress to assess the effects of increased boron doping on device performance.

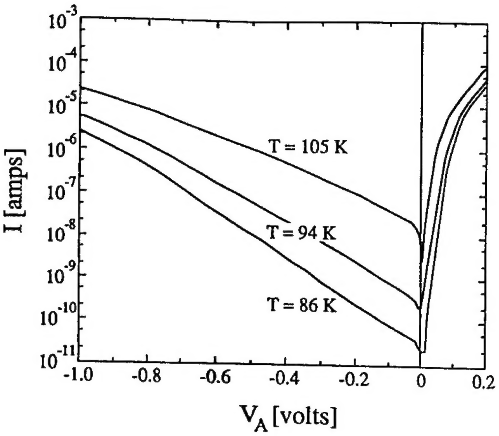

Figure 5. Measured $I(V)$ characteristics for a heterojunction internal photoemission detector ($N_A = 5 \times 10^{19}$ cm^{-3}).

Conclusions

HIP infrared detector structures impose particular demands on the growth process with respect to residual boron doping of silicon layers and morphology of heavily doped Ge_xSi_{1-x} layers. We have shown that minor modifications of our usual procedure allow us to obtain acceptably low residual doping. Thickness undulations in the Ge_xSi_{1-x} absorbing layer are a serious problem for the high boron concentrations required but can be controlled by reducing the growth temperature to 550°C.

We have demonstrated the growth of operating detector structures with barrier heights close to the desired values. Future work will be directed at increasing the quantum efficiency by increasing the doping concentration, together with tailoring of the barrier height to the desired value.

Acknowledgements

The authors gratefully acknowledge support of this work by AFOSR grants F49620-93-1-0387 and F49620-92-J-0155 and the Caltech President's Fund.

References

[1] D A Scribner, M R Kruer, and J M Killiany, *Proc IEEE*, **79**, 66 (1991).
[2] T L Lin and J Maserjian, *Appl Phys Lett*, **57**, 1422 (1990).
[3] H C Liu, L Li, J-M Baribeau, M Buchanan and J G Simmons, *J Appl Phys*, **71**, 2039 (1992).
[4] R A A Kubiak, W Y Leong, M G Dowsett, D S McPhail, R Houghton and E H C Parker, *J Vac Sci Technol.* **A4**, 1905 (1986).
[5] D J Robbins, J L Glasper, A J Pidduck, and A G Cullis, In *Proceedings of 2nd International Conference on Electrical Materials*, p. 477. Materials Research Society, Pittsburgh, PA, USA (1990).
[6] J C Bean, *IEEE Proceedings*, **80**, 556 (1992).
[7] E Kasper and H-J Herzog, *Thin Solid Films*, **44**, 357 (1977).
[8] J C Bean, *Science*, **230** (4722), 127 (1985).

[9]D W Greve, *Mater Sci Engng*, **B18,** 22 (1993).
[10]D W Greve, R Misra, T E Schlesinger, G McLaughlin and M A Capano, *Thin Solid Films*, **222,** 46 (1992).
[11]H Jorke and H Kibbel, *Appl Phys Lett*, **57,** 1763 (1990).
[12]M L Yu, D J Vitkavage and B S Meyerson, *J Appl Phys*, **59,** 4032 (1986).
[13]S S Iyer, S L Delage, and G J Scilla, *Appl. Phys. Lett.*, **52,** 486 (1988).
[14]D W Greve and M Racanelli, *J Electron Mater*, **21,** 593 (1992).
[15]C W Snyder, B G Orr, D Kessler and L M Sander, *Phys Rev Lett*, **66,** 3032 (1991).
[16]A J Pidduck, D J Robbins, A G Cullis, W Y Leong and A M Pitt, *Thin Solid Films*, **222,** 78 (1992).
[17]D J Srolovitz, *Acta Metall*, **37, 621** (1989).
[18]R J Jaccodine, *J. Electrochem Soc*, **110,** 524 (1963).
[19]In Landholt-Börnstein, *Semiconductors*, Vol. 17, pp. 412, 372. Springer-Verlag, Berlin, Germany (1982).
[20]C G Van de Walle and R M Martin, *Phys Rev*, **B34,** 5621 (1986).
[21]T-L Lin, A Ksendzov, S M Dejewski, E W Jones, R W Fathauer, T N Krabach and J Maserjian, *IEEE Trans Electron Devices*, **ED-38,** 1141 (1991).